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THE PARTICLE-MATRIX INTERFACE STRENGTH IN TD-NICHROME.(U)
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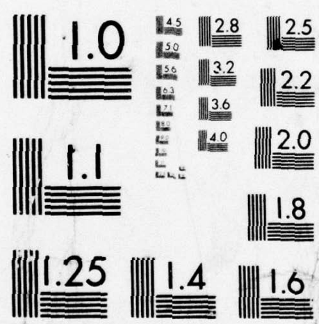

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In TD-Nichrome

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Reprinted from The Microstructure and Design of Alloys
Proceedings of the Third International Conference on the
Strength of Metals and Alloys, Volume 1, August 1973

October 1973

Technical Report No. 4

To the Office of Naval Research

Contract No. N00014-67-A-0117-0011
NR031-689/9-17-70 (471)

Materials Division ✓
Rensselaer Polytechnic Institute
Troy, New York 12181

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THE PARTICLE-MATRIX INTERFACE STRENGTH IN TD-NICHROME

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Introduction

Olsen and Ansell have shown that the yield strength of some two-phase alloys is directly related to the strength of the interface bond between the two phases (1). They found that in TD-Nickel this could result in a tensile yield strength lower than the compressive yield strength, and attributed this strength differential to the separation of the particle-matrix interface in tension at a stress below the normal yield stress. In a later investigation Olsen, Judd, and Ansell measured the stress required to cause separation of the particle-matrix interface, and found that in TD-Nickel the interface strength was indeed low enough to produce an anomolously low tensile yield strength (2).

The present study was undertaken to investigate the effect of a solid solution alloying element on the particle-matrix interface strength in a dispersion strengthened alloy. This effect was related to the tendency for interface decohesion and to the occurrence of a strength differential (SD effect).

Material

The dispersion strengthened alloys used in this investigation were TD-Nickel (thoria particles with a mean diameter of 400 Å and a mean interparticle spacing of 2,000 Å dispersed in a nickel matrix) and TD-Nichrome (thoria particles with a mean diameter of 290 Å and a mean spacing of 830 Å in a 80% nickel - 20% chromium matrix). TD-Nickel and TD-Nichrome (heat 3332) were supplied by Fansteel Corporation in the form of stress relief-annealed sheet. TD-Nichrome (heat 3325) was supplied by Fansteel as 0.250 inch rod in the stress relief only condition and was later annealed at 2400°F.

Experimental Technique

Room temperature tension and compression tests were performed on TD-Nichrome, heat 3325, to determine if this alloy exhibited an SD effect. Tensile specimens were 2.5 inches in length and 0.25 inch in diameter with a gauge length of 0.640 ± 0.002 inch and a reduced section 0.160 ± 0.003 inch in diameter. Compression specimens were machined as cylinders 0.250 ± 0.002 inch long and 0.25 inch in diameter. Both tension and compression tests were performed with a Model TT-CM

Instron Testing machine at a strain rate of 0.0026 $sec.^{-1}$.

Several specimens of TD-Nichrome (heat 3223) and TD-Nickel were cold rolled to 80%, 60%, 40% and 20% of the as-received sheet thickness. The resulting microstructures were examined and compared by transmission electron microscopy.

The particle-matrix interface strength of TD-Nichrome (heat 3323) was measured using the method developed by Olsen et al (2). This is based on the difference in coefficient of thermal expansion between the metal matrix and the ceramic particles. If a specimen is heated extremely rapidly, the metal-oxide system is unable to accommodate the thermal expansion difference and high stresses will develop at the interface. If the change in temperature, ΔT, is large enough, this stress may be sufficient to cause particle-matrix decohesion.

Rapid thermal cycling was performed with the Gleeble, Model 501, described by Savage (3). Sheet specimens were machined to a length of 3.0 in. and width 0.5 in. with a reduced section 1.25 in. long and 0.25 in. wide. Specimens were heated to various temperatures between 650° and 1800°F at a rate of 12,000 to 14,000°F per sec. After holding at temperature for some time, Δt, the samples were tensile tested to fracture at a strain rate of approximately 0.02 $sec.^{-1}$ The fracture surfaces of these high temperature tensile test samples were examined by scanning electron microscopy.

Results

Room temperature mechanical testing of TD-Nichrome revealed that no significant SD effect occurred in this alloy. The average 0.2% yield strengths for the specimens tested were 103,000 psi in tension and 106,000 psi in compression.

Large plastic deformations (cold rolling 40, 60 or 80 per cent) result in the formation of voids at oxide particles in both TD-Nichrome and TD-Nickel. The voids are elongated in the direction of rolling. In TD-Nickel voids are very elongated and appear to be associated with most of the particles. In TD-Nichrome, however, the voids are smaller and less elongated, and fewer particles have voids associated with them.

Fig. 1 shows that after a reduction in thickness of only 20% voids are quite evident in TD-Nickel while few if any voids can be observed in TD-Nichrome.

The results of thermal cycling and high temperature tensile testing are summarized in Fig. 2. The 0.02% yield strengths of specimens tested after holding at 650°, 1200°, 1500°, and 1800°F following rapid heating are plotted as a function of time at temperature (Δt). For the three lower temperatures there is no variation in yield strength with time at temperature. At 1800°F, however, yield strength increases from about 25,000 psi at 0.4 seconds to 32,000 psi for times of 5 seconds or more. Therefore, at some temperature between 1500° and 1800°F a transition occurs from a time independent yield strength to a yield strength which varies with time.

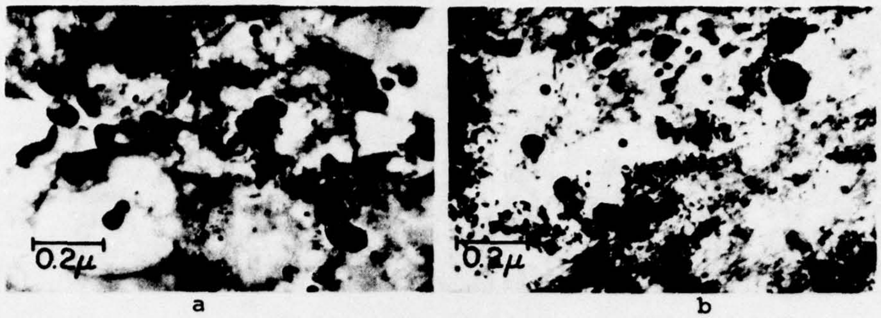


FIG. 1
(a) TD-Nickel sheet, cold rolled 20%.
(b) TD-Nichrome sheet, cold rolled 20%.

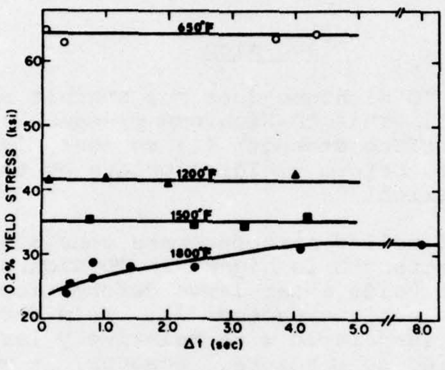
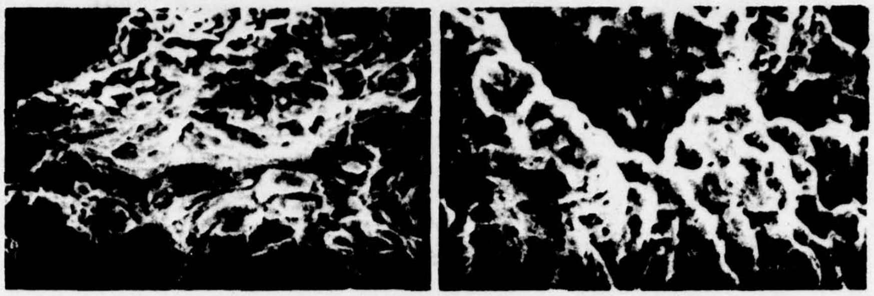


FIG. 2
Yield strength of TD-Nichrome as a function of time at temperature, following rapid heating.

The nature of the transition from time independent to time dependent yield stress was further investigated by studying the high temperature tensile specimen fracture surfaces with the SEM. There is no difference in the appearance of fracture surfaces of specimens tested after various times at temperatures of 1500°F and below. All display the fine dimples associated with ductile fracture in dispersion strengthened alloys.

At 1800°F the fracture surfaces are significantly different. Fig. 3a shows a specimen held for 4.2 seconds at 1800°F prior to loading. The fracture mode in this case appears to be intergranular. Fig. 3b, on the other hand, shows a specimen pulled after 0.8 seconds at 1800°F. Here the facets appear to be rounded and individual grains are not distinguishable. The distinctive feature of this surface is that the facets are marked by many cavities.



a b

FIG. 3
Fracture surfaces of TD-Nichrome pulled to failure after (a) 4.2 sec., (b) 0.8 sec. at 1800°F.

Discussion

It is apparent that TD-Nichrome does not exhibit an SD effect as was observed in TD-Nickel. Thus TD-Nichrome presumably has a relatively high particle-matrix interface strength (1) so that, in tension, decohesion does not occur, before yielding occurs by the same mechanism which operates in compression.

Observations of cold rolled structures are consistant with this view that the interface strength is higher in TD-Nichrome than in TD-Nickel. The formation of voids after large deformations in both TD-Nichrome and TD-Nickel is not unexpected (1). Void formation of this type has most often been associated with relatively large amounts of plastic deformation leading to fracture. However, it was proposed that particle-matrix decohesion could occur at much smaller strains as a result of elastically induced stresses, and lead to macroscopic yielding (1). For very small amounts of plastic deformation voids should be present although they will be so small that they will not be observable by electron microscopy.

Although the structure of the cold rolled specimens does not associate void formation with yielding, it does show that void formation is more prominent in TD-Nickel than in TD-Nichrome, especially for small amounts of deformation. This indicates that interface decohesion occurs at lower stresses in TD-Nickel, and that TD-Nichrome therefore has a stronger particle-matrix interface bond. Therefore, the SD effect would be less in TD-Nichrome than in TD-Nickel, which is in agreement with the results of mechanical testing both in this investigation and as reported by Olsen (4).

A quantitative estimate of the particle-matrix interface strength was obtained using the technique described above (2). The ΔT required to cause interface decohesion is defined by determining the temperature at which the yield strength becomes time dependent. This phenomenon results from the detrimental effect of the presence of voids at the particle-matrix interfaces and the subsequent "healing" of these voids with time at temperature. From this transition temperature the parti-

cle-matrix interface strength can be calculated using the equation of Rao et al (5). The transition temperature, as shown in Fig. 2, is between 1500°F and 1800°F for TD-Nichrome. By calculating the interface stress resulting from heating to each of these temperatures the interface strength is determined to be between 198,000 psi and 238,000 psi. These values are considerably higher than the 40,000 psi measured by Olsen et al for TD-Nickel (2). In TD-Nichrome then it is not expected that interface decohesion will occur prior to yielding, and, therefore the SD effect should not be present.

The SEM studies showed that at temperatures of 1500°F and below, the fracture surfaces were dimpled due to void formation followed by void coalescence and growth. It is believed, however, that voids in these specimens form during plastic deformation and not as a result of rapid heating. No change in yield strength with time at temperature is expected for these alloys if interface decohesion does not occur as a result of rapid heating. Consequently, the appearance of the fracture surface is the same for all times at temperature.

At 1800°F, however, there was a time dependence of the appearance of the fracture surface which accompanied the time dependence of yield strength. For long times at 1800°F the fracture surface has the typical intergranular appearance. There is no evidence of void formation in Fig. 3a. If voids were formed during rapid heating they must have healed before the tensile test was performed. In the samples tested immediately after reaching 1800°F, however, voids have not had sufficient time to heal. The applied stress prevents voids from healing with time and, in fact, enlarges voids so that they can be observed with the SEM. Fig. 3b shows that many cavities remain on the fracture surface and that particles are visible in some of these.

Conclusions

The addition of chromium to the nickel matrix in thoria dispersed alloys increases the particle-matrix interface strength thereby reducing or eliminating the SD-effect. However, rapid heating to 1800°F can produce sufficient stress to cause decohesion of the interface thus decreasing the yield stress for short times at temperature after initial heating.

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Security Classification

DOCUMENT CONTROL DATA - R&D

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1. ORIGINATING ACTIVITY (Corporate author) Rensselaer Polytechnic Institute Troy, New York 12181	2a. REPORT SECURITY CLASSIFICATION Unclassified 2b. GROUP

3. REPORT TITLE

6 The Particle-Matrix Interface Strength in TD-Nichrome.

4. DESCRIPTIVE NOTES (Type of report and inclusive dates)

9 Technical Report.

10 5. AUTHOR(S) (Last name, first name, initial)

J.E. Franklin, ~~J. E.~~ G. Judd, ~~G.~~ and G.S. Ansell ~~G. S.~~

6. REPORT DATE October 1973	7a. TOTAL NO. OF PAGES 6	7b. NO. OF REFS 5

8a. CONTRACT OR GRANT NO. 15 N00014-67-A-0117-0011	9a. ORIGINATOR'S REPORT NUMBER(S) Technical Report No. 4 14 TR-4
b. PROJECT NO. c. NR031-689/9-17-70(471) 16 NR-031-689	9b. OTHER REPORT NO(S) (Any other numbers that may be assigned this report) 11 Oct 73 12 8p.

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The particle-matrix interface strength between the thoria dispersed phase and the nichrome matrix was measured by means of a thermal shock method. The interface strength was determined to be between 198,000 and 238,000 psi, considerably higher than that found previously for TD-Nickel, 40,000 psi. These results were then correlated with the formation of voids during cold rolling of the alloys.

DD FORM 1 JAN 64 1473

Unclassified
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14. KEY WORDS	LINK A		LINK B		LINK C	
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